VOL. 6. AUGUST, 1952 No. 4

WELDING RESEARCH

Published by
BRITISH WELDING RESEARCH ASSOCIATION · 29 PARK CRESCENT · LONDON · W.1

WELDING RESEARCH

VOL. 6 NO. 4 AUGUST, 1952

Published by

BRITISH WELDING RESEARCH ASSOCIATION

Head Office and Metallurgical Laboratory

29, PARK CRESCENT, LONDON, W.1

Telephone: Langham 7485/9

Telegrams: Weldaserch, Wesdo, London

Research Station

Abington Hall, Abington, nr. Cambridge

Telephone: Linton 375

Telegrams: Weldaserch, Cambridge

Director of Research

H. G. TAYLOR,
D.SC.(ENG.), D.I.C., M.I.E.E., F.INST.P.

Secretary

A. O'NEILL, B.SC.(ECON.), A.C.I.S.

Editor and Publications Officer

J. H. GAMESON

CONTENTS

	Page
Association Notes	68*r*
Experiments on Brittle Fracture of Steel – resulting from residual welding stresses ..	70*r*
Argon-arc Welding of Copper	83*r*

ASSOCIATION NOTES

THE MECHANICS OF NOTCH BRITTLE FRACTURE

A new approach has been made to the solution of this very difficult and important problem by Dr. A. A. Wells of the British Welding Research Association, and a brief account of the work done is given below. A more complete report of the work on this subject is being prepared and should be available soon for publication in this journal.

Notch brittleness, as a property of mild steels at temperatures no lower than 0° C, has been known for many years. The sudden type of failure, as a crack emanating from the notch, accompanied by no recognisable ductile extension to give prior warning, may be catastrophic in its effect. The fractured surface is recognised afterwards by its crystalline appearance, as distinct from the fibrous surface which is the final result from an extension with full ductility. Since notches in structures can seldom be entirely removed, much research is being conducted with the object of ensuring that steels are not used which show notch brittleness at the lowest service temperatures. It is known, however, that structures have been successfully used where notches are present, and where the material is in the notch brittle condition. Moreover, for reasons of economy, it may not always be possible in the future to eliminate from use all notch brittle steels. It is therefore important to establish the conditions of stress, both residual and from external loads, at which a notch brittle failure can occur, and for this reason the B.W.R.A. has undertaken a study of the mechanics of notch brittle fracture. For the time being, the study is limited to the case of static loads applied to mild steel just below the transition temperature at which notch brittleness is first observed. Only the main conclusions are stated here, and the fuller discussion, on which the results are based, will be published at a later date.

Not even a crack in a fully brittle material can extend without the expenditure of some energy in creating the freshly cracked surface. Griffith* showed that the work is done against a surface tension, just as for liquids. He also showed that the energy of propagation could come from the elastic strain energy stored in the body under stress, and developed an expression for the critical stress necessary to propagate a crack of given initial length, assuming the crack to be perpendicular to the direction of tension. This he did by equating the surface tension for surface energy as it becomes when related to unit length of crack) to the fall of stored elastic strain energy in the plate, having its ends clamped at a given extension, when the crack was extended by unit distance. The derived expression fitted well the experimental results for glass, for which the surface tension could be determined by extrapolation from the liquid phase. Now, tensile tests of notched mild steel plates show brittle breaking loads which, if the Griffith formula were to be applied, would call for surface tensions 10,000 times greater than the independently measured values. This might be caused by the addition of localised plastic flow occurring along the propagating crack. If it were so, B.W.R.A. theoretical studies of heat flow in resistance welding showed that temperature waves, of amplitude about 1° C, would move outwards from the cracked surfaces at the instant of fracture, since heat would be generated from the plastic flow. Thermocouple measurements confirmed the existence of these temperature waves, as shown by the oscillograph record in fig. 1. From their amplitudes, apparent surface energies were readily calculated, to be compared with measured breaking loads. For a bar with deep and sharp symmetrical external notches, the Griffith formula for the breaking load P on unit thickness of plate, was found to be:—

$$P = \sqrt{\pi ESa}$$

where E is Young's modulus, S is the surface energy, and 2a is the width of the plate at the narrowest notched cross-section. Measured breaking loads of fifteen specimens of notch brittle steel sharply notched to different depths, when plotted against $\sqrt{2aS}$, as also derived from measurements, compare well with the theoretical line, calculated from the formula above, when a value of 13,500 tons/sq. in. for E is used. (See fig. 2). Exceptions occur at the extreme breaking loads, where ductile fractures took place.

The experiments clearly showed that the energy necessary to propagate the brittle fractures came entirely from the stored elastic strain energy in the plates, and not from the external loads. The ability of residual stresses to contribute energy to the propagation of brittle cracks was thus demonstrated. It was also clear, however, that the apparent surface energy was not a constant for the material, as in Griffith's case of a truly brittle material. Another criterion had therefore to be sought for the initiation of brittle fractures.

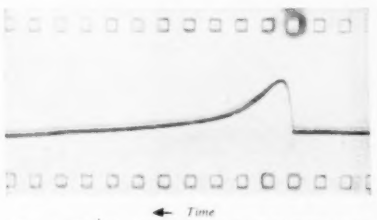


Fig. 1. Recorded temperature wave adjacent to a brittle fracture.

Evidence for the acceptable criterion of initiation was eventually found from the results of static tensile tests

* GRIFFITH, A. A., Phil. Trans. Roy. Soc., Series A, Vol. 221, 1921, p. 163. Also see Proc. Intern. Cong. App. Mech., Delft, 1924, p. 55.

on notched plates, from 1" to 72" wide, published by seven investigators in the U.S.A. and Great Britain. The notches in these tests ranged from plain sawcuts to deliberately introduced fatigue cracks. It was found that in no case was the breaking stress on the minimum notched cross sectional area less than the yield point of the material. The lowest breaking stresses were for those specimens whose width was large compared with their thickness; other specimens, of width/thickness ratio approaching unity, and above, showed average breaking stresses up to double the yield point.

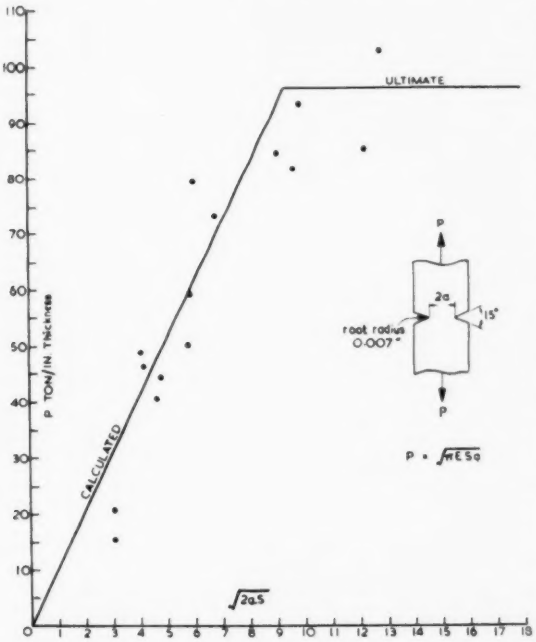


Fig. 2. Relation between measured surface energy and breaking loads for externally notched specimens.

Comparison of these results with those from experimental and theoretical studies of plastic flow in notched bars then showed that these were the stresses at which a plastic zone would first extend, from the notch root, right across the notched area, and plastic extension of the plate as a whole would be possible. The thin plate case corresponded with plane stress, when no component of stress could exist across the thickness, and the thick plate case represented plane strain, when no deformation could exist across the thickness. For the latter case, general yield would be predicted at an average stress on the notched area of $(1 + \frac{\pi}{2})$ times the yield point, this being defined as in the usual tensile test.

The significant feature of general yield is that it is the point at which large plastic strains are first allowed to occur at the root of the notch. While a continuous elastic zone still exists from top to bottom of the plate, general yield and extensive plastic flow at the root of the notch are both inhibited. This point was readily demonstrated by means of a small gauge length extensometer placed near the root of a notch, on a notched plate for which a load extension curve was also determined.

Now, the true stress at the instant of failure of an ordinary tensile specimen of mild steel of usual ductility is always three or more times the yield stress, and yet this is insufficient to cause transcrystalline, or brittle, fracture. Moreover, the plasticity studies preclude the possibility of existence of any stress greater than $(1 + \frac{\pi}{2})$ times the yield, when general yield of a notched specimen occurs without simultaneous strain hardening.

The conclusion from these several observations is that strain hardening at the root of the notch, following the considerable increase of strain at the onset of general yield, is an important mechanism by which the stress is allowed to concentrate to a value sufficient to start brittle failure. Whether the brittle failure then continues, depends on the availability of elastic strain energy to maintain the process along the advancing crack.

It would be unwise, immediately to generalise these results to complicated structures, and those containing welds, with resultant extensive residual stress systems not eliminated by stress relief. In spite of this, the results are reassuring in special cases at atmospheric temperatures, and with the mild steels showing usual ductility in the tensile test. Where stress relief has been complete, it would appear that external loads giving rise to average stresses exceeding the yield point are necessary before brittle failure can take place, even if notches are present. Commercial structures are usually designed to avoid stresses of this magnitude in any case, so that safety ought to be assured by adequate stress relief. Conversely, it becomes possible to understand why brittle fractures may occur, even where external loads are absent, since it is well known that the reaction stresses due to longitudinal and transverse contractions at fusion welds frequently exceed yield point magnitude. The formation of sufficiently sharp notches or natural cracks to cause brittle factures at lower than yield point average stresses, with static loads, may be possible, but has certainly yet to be demonstrated.

R.83

EXPERIMENTS ON BRITTLE FRACTURE OF STEEL — RESULTING FROM RESIDUAL WELDING STRESSES

The following report of work done on behalf of the FE.19 Committee of the British Welding Research Association is being made generally available because of the importance of the results achieved. This report is a preliminary statement of the experimental results, and further work is clearly to be envisaged before the full implications can be said to have been adequately proved experimentally. The Committee, and particularly the author, would wish this situation to be understood.

By R. Weck, Ph.D.

SUMMARY

The report describes a series of five tests in which butt welds were made in $\frac{1}{2}$ in. thick plates under conditions of very severe restraint, in order to produce very high and complex residual stresses. The welded assemblies were slowly cooled to room temperature; cooling was then continued in a special cooling chamber to temperatures in the vicinity of the transition temperature ranges of the materials used. The steels were:

(i) ordinary coarse grained mild steel with a high Izod transition temperature;
(ii) a normalised low-carbon, open hearth, aluminium-killed, fine-grained steel with a low Izod transition temperature;
(iii) a normalised, weldable, high tensile, chromium-molybdenum steel of fine grain size and low Izod transition temperature.

A notch was provided in the specimens consisting of a drilled hole extended by saw cuts parallel to the butt weld.

It was found that the residual stresses introduced in the test piece led to spontaneous brittle fracture in the coarse grained steel at a temperature below the transition range of the material. The transition range was determined by Izod tests though it was not sharply defined. In the fine grained materials the residual stresses did not produce brittle fracture, even when the specimens were severely hammered at very low temperatures, in the one case below, and in the other case within the clearly defined transition range, as determined by Izod tests.

The conclusion is drawn that residual stresses may produce spontaneous brittle fracture in the presence of notches only at temperatures below the transition temperature of the material, but that even below the transition temperature of the material and in the presence of notches, will not inevitably produce fracture.

INTRODUCTION

The term "Brittle Fracture" has been much used in recent years in the more restricted sense of denoting the apparently anomalous type of fracture experienced with steels whose ductility, as determined by the standard tensile test carried out at room temperature, is perfectly satisfactory, but which may under certain circumstances, break without showing any obvious signs of deformation. This type of fracture usually takes place at lower temperatures, under static or impact loading and starts from a notch or other discontinuity. The speed of propagation of the fracture is high, so that it appears to take place suddenly. The fractured surfaces exhibit a typical, easily recognisable, coarsely crystalline appearance which should not be mistaken, though it sometimes is, for fatigue failure, with which it has nothing in common except the absence of any visible signs of deformation.

Such brittle fractures have been experienced with welded steel structures more frequently than with riveted structures. One common characteristic of these failures is that they appear to have occurred spontaneously in the sense that the structure was carrying either normal service loading or even only a small part of the design load. The Hasselt and Herrenthal-Olsen bridges in Belgium and the Hardenberg-strasse and Rudersdorf bridges in Germany, for instance broke when the bridges carried no traffic; the American T.2 Tankers S.S. *Schenectady* and *Esso Manhatten* broke in two, the one in still water when subject to a maximum calculated stress of 4·8 tons per sq.in., and the other at sea when subject to a calculated stress of 5·5 tons per sq.in.

It has been known that welding may introduce very high residual stresses. It was natural therefore that these failures in welded structures should have been attributed to residual stresses. It remained to be explained, however, why the fractures, even if they were due to residual stresses, should be brittle in what otherwise appeared to be a perfectly ductile material.

There is one school of thought who believe that it is a peculiar property of residual stresses to produce brittle fracture in ductile materials. It is agreed that residual stresses may be equi-triaxial at certain points and that plastic deformations cannot occur there because the octahedral shear stress would be too small to produce plastic deformations before the fracture strength of the material was exceeded

by the principal normal stress. Whilst it cannot be denied that residual stress systems of this type could occur in consequence of welding, it is equally undeniable that such stress systems must needs be confined to comparatively small areas—for instance, to the immediate vicinity of fillet welded Tee connections. This view may explain the occurrence of comparatively small localised cracks, but not the extensive fractures which have certainly passed through material where such triaxial stress systems could not be supposed to have been present by any stretch of imagination.

There is another school of thought claiming that the properties of the material are changed by the presence of complex residual stress systems, such as are created by welding, so that residual stresses embrittle an otherwise ductile material.

During recent years it was recognised that brittle fracture in steel can be produced by static external loading, provided that a notch is present and that the temperature of the material is lowered to below what is called the transition range of the steel, where this transition range again depends not only on the material and its thickness, but also on the severity of the notch and the straining speed.

The demonstration of this fact at any rate disposed of the widely held belief that brittle fracture was a peculiar effect of residual stresses only. At the same time there remained a number of questions to be answered:

(i) Could residual stresses of such complexity and severity as are likely to occur in welded structures result in brittle fracture, at temperatures above or below the transition range as determined by some form of notch test, with or without external load,
  (*a*) without the presence of a notch?
  (*b*) in the presence of a notch?

(ii) Will the presence of residual stresses facilitate brittle failure of any given assembly under external load, given that an identical assembly free from residual stresses would fail in a ductile manner when subject to the same load and when tested at the same temperature?

(iii) Is the maximum load carrying capacity of a structure reduced by the presence of residual stresses?

All these questions are, of course, related. The last one was discussed elsewhere[1] by one of the authors, where it was shown that as long as the material was sufficiently ductile to fail by plastic deformation, residual stresses could not affect the ultimate load carrying capacity of a structure because residual stresses and load stresses could not be additive everywhere. If there are regions of reduced carrying capacity because the residual stresses are augmented by the service stresses, there must necessarily be regions where the service stresses are of opposite sign to the residual stresses and where, therefore, the carrying capacity is increased by the same amount. This conclusion is in entire agreement with all known experimental facts. The carrying capacity of a structure could be affected, however, if residual stresses could in some way prevent the ductile behaviour of the material, because then the above argument on superposition does not necessarily apply. Hence the last question, therefore, is reduced to this:

(iv) Have residual stresses an embrittling effect on the material?

This question is meaningless in the light of what is known now on brittle fracture if it does not mean that residual stresses either shift the transition range towards higher temperatures or else that a less severe notch, which would ordinarily result in ductile fracture at a certain temperature would suffice to produce brittle fracture at the same temperature in the presence of residual stresses.

In a previous series of experiments[2] it was found that residual stresses equal to the yield point of the material must be expected to arise. Even in the case of a simple butt weld made under quite moderate restraint, stresses of this magnitude can be shown to exist. It can be concluded from these experiments that it was merely a question of refinement of measuring methods whether these stresses could be discovered or not in any particular case. If an external load is superimposed the stresses at the points where they are already equal to the yield point of the material are not increased, because in the absence of notches plastic deformation sets in immediately on loading, even at temperatures some 40 deg. C. below the transition range of the steel as determined by Tipper tests.[3] It was shown that residual stresses equal to the yield point of the material could not produce brittle fracture in the absence of notches. A superimposed external load would have to reach a value equal to the cross-sectional area of the plate multiplied by the yield point of the material before the yield point was reached throughout the cross section, despite the fact that stresses up to the yield point were present, before any external load was applied.

These experiments answered the question postulated under (i) (*a*) in the sense that residual stresses could not produce brittle fracture with or without an additional external load in the absence of notch effects and discontinuities, even at temperatures well below the transition range of the steel as determined by notch-tensile tests.

Whilst the residual stresses in the specimens used in the previous experiments were as high as they could possibly be, the stresses, however, were only biaxial and not of such complexity as might occur in a more intricate structure. There was also the fact that the specimens contained no notches. It was clearly desirable to continue these experiments in a much more severe form by using specimens containing notches and of such construction that the state of stress should not only be at least as severe as that realised in previous specimens, but should if possible be far more complex. The outcome of these considerations were the tests described in this report which are, therefore, a logical continuation of the previous experiments. It was expected that the new series of tests would at least give valuable pointers towards the solution of the problems postulated above.

Three different steels of greatly differing transition ranges were used in the present series of experiments. Plates were joined by butt welds under such severe restraint as would rarely be experienced in welded construction and which was certain to result in residual stresses at least as high as those measured in the previous experiments. The specimens were notched and cooled to very low temperatures in the expectation that some of them would, and some of them would not fail spontaneously under the influence of residual stresses alone. In this way it was hoped to simulate on a very small scale the catastrophic spontaneous fractures experienced in large structures, and it was expected that from these experiments an approximate assessment could be made of the part played in these failures by the low temperature notch sensitivity of the steel on the one hand and the residual stresses on the other.

TESTING PROCEDURE

The specimens consisted of two $\frac{3}{4}$-in. thick butt welded plates which, prior to making the butt weld, were securely welded to a very rigid restraining frame in order to produce very high residual stresses. This restraining frame was fabricated from machined 8 in. by 4 in. steel billets and stress relieved for twelve hours at 680 deg. C. The frame consisted of two beams of 8 in. by 4 in. solid section and two columns of 6 in. by 4 in. solid section as shown in Figs. 1(*a*) and 1(*b*).

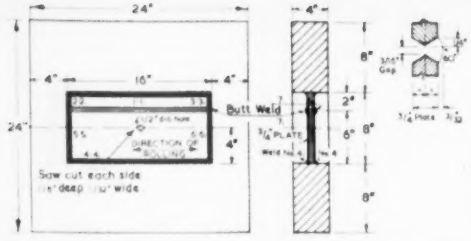


Fig. 1a. Test piece for restrained butt weld without central rib. Test 1.

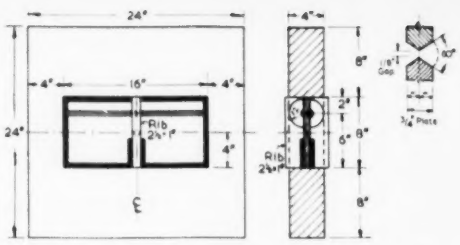


Fig. 1b. Test piece for restrained butt weld with central rib. Tests 2-5.

BUTT WELDS. (*Test* 1, *electrode* "*A*," 200 *amps.*) *All other tests with electrode* "*B*."
OTHER WELDS. Electrode "*B*," $\frac{3}{16}$ *in. at* 150 *amps. and* $\frac{1}{4}$ *in. at* 250 *amps.*
WELDING SEQUENCE. As shown—$1-1_1$, $2-2_1$, $3-3_1$, $4-4_1$, $5-5_1$, $6-6_1$, *allow work to cool to room temperature (screened from draughts), then weld butt* $(7-7_1)$ *and allow to cool.*

Two pieces of $\frac{3}{4}$-in. plate, one 16 in. by 2 in. and the other one 16 in. by 6 in., were welded into the opening by heavy fillet welds and in the sequence shown in Fig. 1(*b*). The larger of the two plates was notched by drilling a $\frac{1}{2}$ in. diameter hole in the centre which was extended laterally, parallel to the butt weld, by two saw cuts $\frac{1}{8}$ in. deep and $\frac{1}{32}$ in. wide.

The butt weld was made at atmospheric temperature in the case of Materials 1 and 2 after each of the two pieces of plate had been welded along the edges to the frame and the whole test piece had cooled to room temperature. In order to avoid the occurrence of hardened zones near the weld, all welding of Material 3 for specimen No. 5 was carried out at 190 deg. C. preheat. When the butt weld cools it contracts and reaction stresses develop in the plates as a consequence of the contraction. The reaction stresses produce stress concentrations at the bottom of the saw cut. The plate material was used in such a way that reaction stresses developed at right angles to the direction of rolling, that is, in the most unfavourable direction.

If this test set-up is compared with that used in the investigation previously mentioned, it will be seen that the restraint is much more severe and that, therefore, the stresses resulting from welding must have been at least as severe as those measured in the previous investigation which were found to be at the yield point of the material.

It could not be said, however, that there were in this test any extensive areas where the residual stresses would be triaxial other than in the immediate vicinity of the fillet welds by means of which the specimens were attached to the frame.

Five tests were carried out, one without central stiffeners and four with central stiffeners. In the latter case the conditions of restraint and presumably the state of stress were, therefore, made more complex by welding in on each side of the plate a central stiffener 1 in. thick and $2\frac{1}{2}$ in. deep as shown in Fig. 1(*b*). A $1\frac{1}{2}$ in. diameter semi-circular hole was provided in the stiffener in order to facilitate the making of the butt weld. This stiffener was welded by fillet welds to the frame and to the plate, prior to making the butt weld, to within approximately 1 in. of the notch in the plate. The fillet welds were not extended right up to the notch so as to avoid setting up initial compressive stresses at the edges of the saw cut before the butt weld was made. By stopping the fillet welds of the stiffener in front of the hole, tensile stresses would occur at the bottom of the saw cut, before the butt weld was made. The stiffener also effectively shortens the "free" length of the specimen so that the butt welds, after these stiffeners have been welded in, will be under very severe restraint. There would also be triaxial stresses present in the vicinity of the fillet welds connecting the stiffeners.

When welding of the specimens was completed and the whole assembly had reached room temperature, it was placed into a cooling chamber consisting of a double-walled wooden box with slag wool between the walls for insulation and with a metal trough at one end to contain the cooling mixture. A small fan was arranged at one end of the trough to circulate the cooling gas. The arrangement is shown in Fig. 2. The restraining frame and welded plate were placed in this

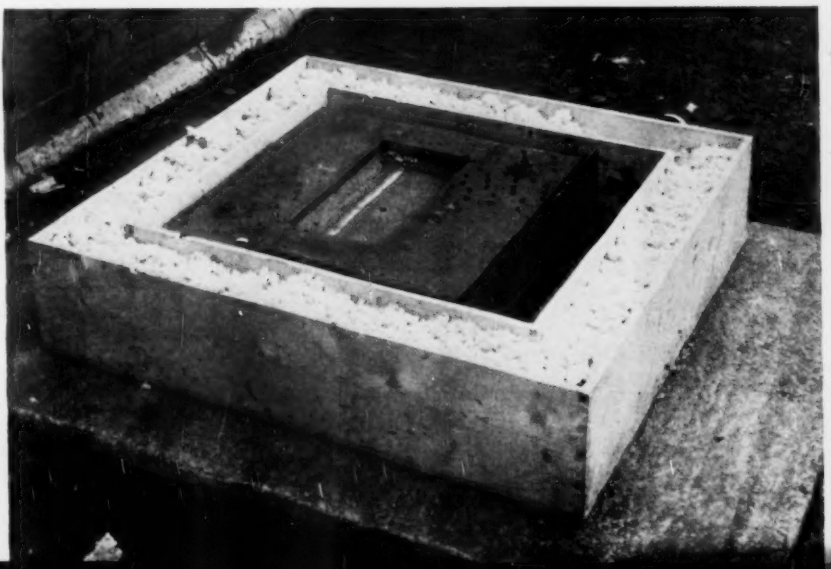


Fig. 2 (*right*). First type of testing plate welded in heavy frame in insulated cooling chamber.

cooling chamber spaced on battens so as to be clear of the insulated box, leaving a gap to allow free access of the cooling gases. Solid CO_2 was put in the trough prior to putting on the insulated lid and liquid oxygen was poured on top of the CO_2 through a funnel passing through the lid of the box as shown in Fig. 3. By switching on the fan the cooling gases were circulated round the heavy restraining frame. This arrangement enabled the whole assembly to be cooled down very slowly at a controlled rate. Thermocouples were arranged to contact insulated positions in the heavy restraining frame and in the $\frac{1}{4}$-in. plate at the position of the drilled hole with the saw slots. A gramophone pick-up provided with a long needle connected to the weld, and an amplifier and loud speaker were arranged so that any noise indicating the occurrence of a crack could be easily detected. This was not used after the first test because the report accompanying the crack was audible without the pick-up and loud speaker.

Some difficulty was experienced in making the butt weld with the first electrode that was tried, called electrode A in this report. The butt weld cracked with several hot cracks and for the second and further specimens another electrode, called electrode B in this report, was used. With this electrode no serious hot cracking trouble was experienced. The hot cracks in the first weld were rewelded and no further hot cracks were seen on the surface after this had been done, although after the specimen was cut up some hot cracks were found as is shown in Fig. 4. Hot cracks were also found in the specimen of Test 4, which was welded with electrode B as is shown in Fig. 5. No hot cracks were found in Test 5 which was welded with preheat as can be seen from the macro section in Fig. 6.

Fig. 3. Cooling chamber in operation, showing sensitive pick-up with probe touching test specimen.

MATERIAL USED IN THE TESTS

The tests were carried out with three different materials. Their chemical analyses are given in Table I and the mechanical properties for the materials in the condition in which they were used are given in Table II.

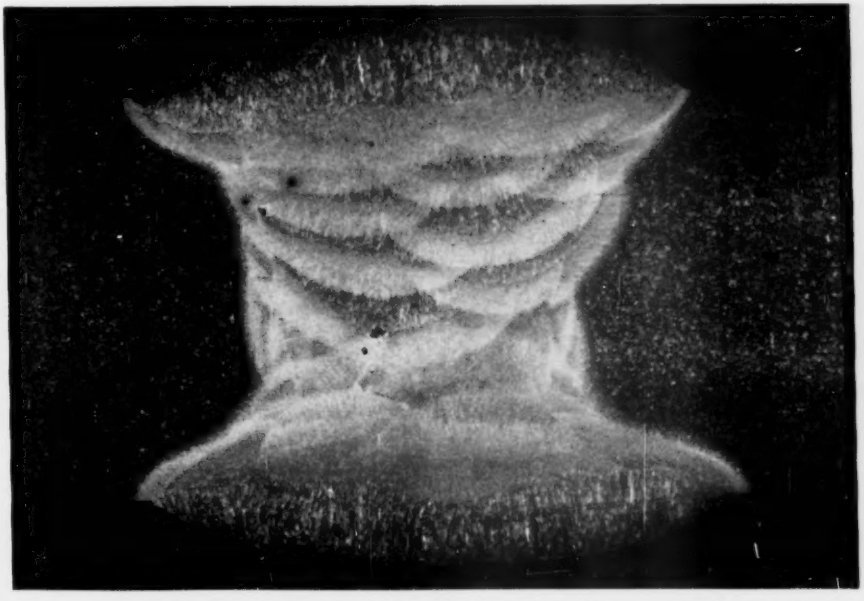


Fig. 4. Photomacrograph of main butt weld. Material No. 1, Test 1. Note presence of cracks in weld.

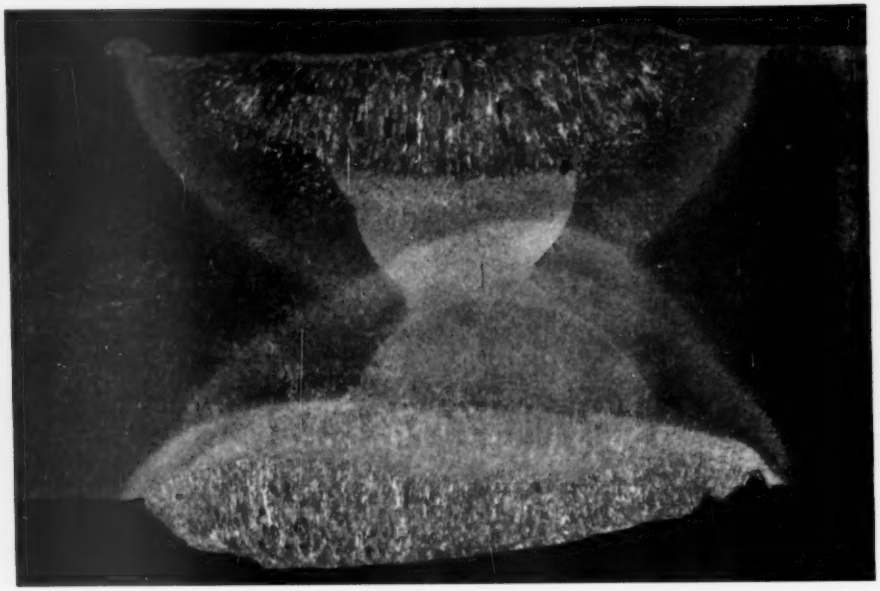


Fig. 5. Photomacrograph of main butt weld, Material No. 2, Test 4. Note hot crack at centre of weld.

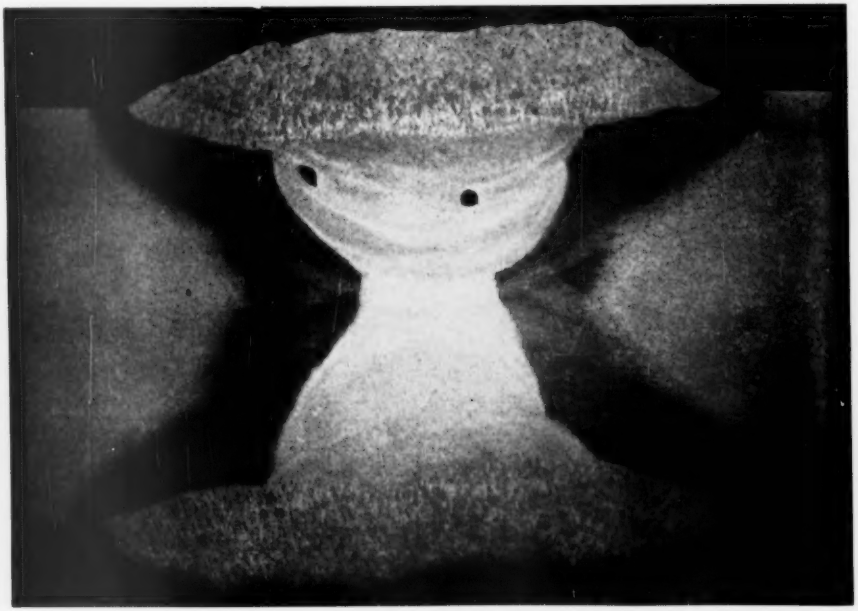


Fig. 6. Photomacrograph of butt weld, Material No. 3, Test 5, welded with 190 deg. C. preheat. No hot cracks.

Grain Size

MATERIAL 1 was ¾ in. thick plate material drawn from stock. It had a coarse McQuaid-Ehn Grain Size greater than 1. A photomicrograph of the material in the as-rolled condition in which it was used for the tests is shown in Fig. 7.

MATERIAL 2 was a low-carbon, open hearth, aluminium-killed steel. The McQuaid-Ehn Grain Size was fine, 9 to 6, and the material was used in the normalised condition as shown in the micrograph, Fig. 8.

MATERIAL 3 was a high yield point, weldable, chromium-molybdenum steel. The McQuaid-Ehn Grain Size of this material was also fine, 8 to 9, some 7. The material was used in the normalised condition as shown in the micrograph, Fig. 9.

Izod Transition Temperatures

Curves for the variation of Izod impact values with temperature for the three materials were obtained on standard 10 mm. by 10 mm. Izod specimens cut longitudinally with the grain of the plates. These are reproduced in Figs. 10, 11 and 12.

The stock plate material, Material 1, shows a gradual drop in Izod value between + 30 deg. C. and − 20 deg. C. without a definite transition.

The low carbon, open hearth, aluminium-killed steel, Material 2, shows a high value down to a temperature of − 40 deg. C. A marked drop then occurs and low Izod values are reached for − 50 deg. C. The transition range of this steel therefore lies between − 40 deg. C. and − 50 deg. C.

The weldable, high yield point, chromium-molybdenum steel, Material 3, shows high Izod values down to − 60 deg. C., the transition range for this steel lying between − 60 deg. and − 90 deg. C.

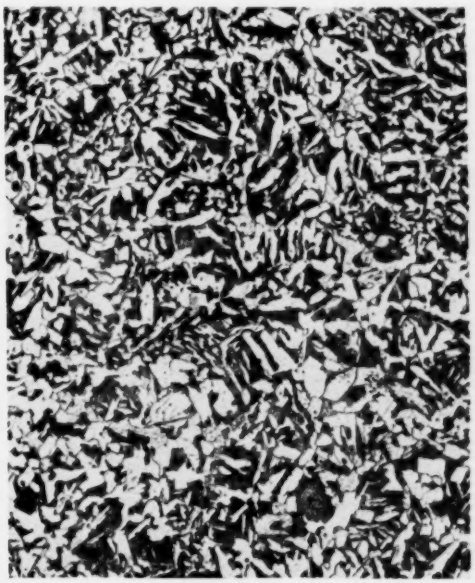
It would have been desirable to determine the transition temperature ranges of the three materials also by means of notched tensile tests. This, however, was not carried out as it would have required a series of notched tensile tests at a number of different temperatures for each of the three materials.

Table I. — Chemical Analyses of Materials

	MATERIAL 1 (per cent.)	MATERIAL 2 (per cent.)	MATERIAL 3 (per cent.)
Phosphorus	0·046	0·018	0·030
Sulphur	0·049	0·025	0·034
Silicon	0·091	0·053	0·181
Manganese	0·66	0·455	0·692
Carbon	0·274	0·132	0·229
Copper	0·04	0·156	0·062
Nickel	0·078	0·113	0·067
Chromium	Nil	0·061	0·759
Molybdenum	Nil	Nil	0·255
Arsenic	0·054	0·043	0·045
Tin	0·004	0·017	0·022
Nitrogen	0·006	0·0065	0·0065
Aluminium	Trace	0·024	0·014
Alumina	Trace	0·012	0·009

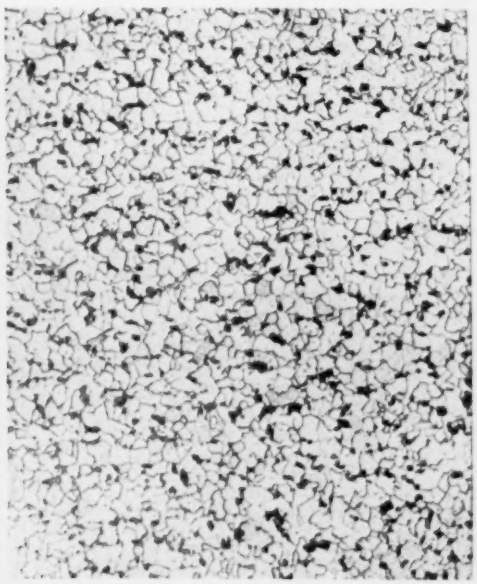
Table II. — Mechanical Properties of Materials

Mechanical Properties	MATERIAL 1	MATERIAL 2	MATERIAL 3
Limit of proportionality (tons/sq.in.)	14·8	—	—
Yield point (tons/sq.in.)	—	17·3	—
0·1 per cent. proof stress (tons/sq.in.)	16·6	—	—
0·2 per cent. proof stress (tons/sq. in.)	16·8	—	—
0·5 per cent. proof stress (tons/sq.in.)	18·0	—	19·8
Ultimate tensile strength (tons/sq. in.)	33·1	25·7	35·0
Elongation $4\sqrt{A}$ (per cent.)	32·5	36·3	36·5
Reduction of Area (per cent.)	57·3	69·0	69·0
Vickers hardness	164	130	172
Izod at 20 deg. C. (ft.-lb.)	30	90	95



×130

Fig. 7. Photomicrograph of ¾-in. plate (Material No. 1.)



×100

Fig. 8. Photomicrograph of low-carbon open hearth aluminium-killed fine grain steel. (Material No. 2).

100

Fig. 9. Photomicrograph of weldable high tensile chromium-molybdenum steel. (Material No. 3)

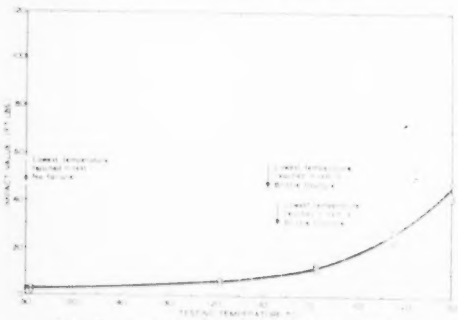


Fig. 10. Izod values for ¾-in. plate (as-received). Material No. 1.

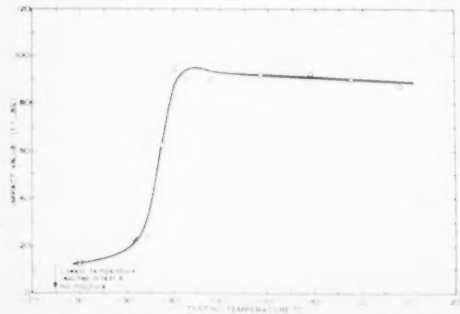


Fig. 11. Izod values for the low-carbon open-hearth killed steel (normalised). Material No. 2.

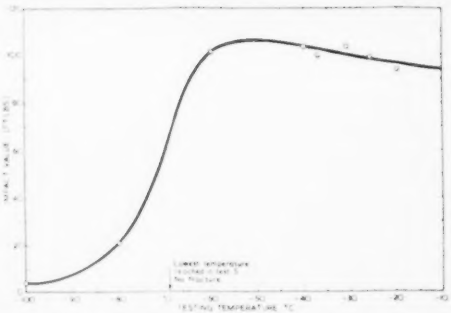


Fig. 12. Izod values for the high yield point, Cr-Mo steel (normalised) Material No. 3.

Notched Tensile Tests

One notched tensile test, however, was carried out with each of the three materials in the following manner:

> Test specimens 1½ in. wide by ¾ in. thick were machined from the ¾ in. plate material. A sharp 45 deg. transverse notch was milled in one edge, the specimen was placed in a 50 ton tensile testing machine; a thermo-couple was placed in the notch, the leads being carried in a groove milled along one of the edges. This groove can be seen in Fig. 13, which shows a specimen cut from the stock plate material after fracture at —11 deg. C. A cooling bath as shown in Fig. 14 was arranged to surround the notched area, so that the material could be cooled for a length of 2 in. above and below the notch. A load was applied to the specimen at room temperature, approximately 1 ton per sq.in. below the yield point as found in the standard tensile test recorded above. The cooling mixture was then introduced into the cooling trough, which slowly cooled the specimen to the following temperatures:
>
> | | | | |
> |---|---|---|---|
> | Material 1 | .. | .. | —11 deg. C. |
> | Material 2 | .. | .. | —43 deg. C. |
> | Material 3 | .. | .. | —64 deg. C. |
>
> At these temperatures no fracture occurred in any of the specimens.

In the case of Material 1 the load was gradually increased to give a stress of 26·1 tons per sq.in. on the net area of the cross section keeping the temperature constant at —11 deg. C., when fracture suddenly occurred with very little plastic deformation at the root of the notch, otherwise the fracture was 100 per cent. crystalline, as shown in Fig. 13.

For Material 2 the load was gradually increased until the stress for the net cross-section reached 26·5 tons per sq.in. keeping the temperature constant at —43 deg. C., when the specimen gradually tore across from the notch giving a 100 per cent. ductile fracture, as shown in Fig. 15 (lower half).

For Material 3 the load was gradually increased until the stress at the net cross-section reached 26·2 tons per sq.in. whilst the temperature was kept constant at —64 deg. C. At this temperature the specimen tore slowly starting from the notch, resulting in a 100 per cent. ductile fracture, as shown in Fig. 15 (upper half).

From these notched tensile tests one may conclude that for Material 1 the transition temperature as determined by the notched tensile test lies above —11 deg. C. and for

Fig. 13 (*below*). Notch tensile test piece of Material No. 1 after test with groove along edge for thermocouple wires. Brittle fracture at −11 deg. C.

Fig. 14 (*above*). Notched tensile test piece in testing machine. Notched area surrounded by cooling bath.

Fig. 15 (*below*). Notched tensile test pieces of Material No. 2 (*lower*) and Material No. 3 (*upper*) after test. Ductile failures at −43 deg. C. and −64 deg. C. respectively.

Materials 2 and 3 it lies below −43 deg. C. and −64 deg. C. respectively. It is likely that these figures would be somewhat higher if the full plate thickness had been used in the notched tensile tests.

RESULTS OF EXPERIMENTS

First Test—Material 1

This test piece, made from coarse grained steel and without stiffeners, is shown in Fig. 2 placed in the cooling chamber. It was inserted into the cooling chamber where it was slowly cooled to a temperature of −60 deg. C., which was very far below the transition range of the material. The reduction in temperature was very gradual so that no thermal stresses, in addition to the welding stresses, should have arisen during cooling because of differences in temperature between frame and plate. The specimen was held at −60 deg. C. for about one hour but no cracks occurred.

Second Test—Material 1

This test piece, as all the following test pieces, was reinforced with stiffeners as shown in Fig. 16 and presumably contained more severe and complex residual stresses than the first specimen. One hour after the butt weld was completed the specimen was cooled slowly in the same manner as specimen 1. Sudden brittle fracture of the specimen occurred spontaneously with a loud report, starting from the saw slots at a temperature of −9 deg. C. as measured by a shielded thermocouple placed in the hole in the stiffener. Because of the larger mass of the frame, this had only cooled to a temperature of −6 deg. C., so that there was a temperature difference between plate and frame of 3 deg. C. The plate if free would have contracted somewhat more than the frame. Since the plate was restrained it could not contract more than the frame, with the result that the reaction stresses in the plate were slightly increased during cooling. The stresses arising in consequence of this temperature difference are biaxial and can be calculated as follows:

The strains e_x and e_y in the x and y directions in consequence of the temperature difference T between frame and plate are given by

$$e_{T,x} = e_{T,y} = \alpha T$$

where α is the linear coefficient of expansion. The stress components $R_{T,x}$ and $R_{T,y}$, due to the temperature difference, are accompanied by strains:

$$e_{0,x} = \frac{1}{E}\left(R_x - \frac{1}{m}R_y\right)$$

$$e_{0,y} = \frac{1}{E}\left(R_y - \frac{1}{m}R_x\right)$$

On the assumption that the frame is perfectly rigid and cannot deform under the influence of the stresses R_x and R_y, the strains due to temperature change must be equal to those due to stress:

$$e_{0,x} = e_{T,x}$$

$$e_{0,y} = e_{T,y}$$

Hence

$$R_x - \frac{1}{m}R_y = E\alpha T$$

$$R_y - \frac{1}{m}R_x = E\alpha T$$

and

$$R_x = R_y = E\alpha T\frac{m}{m-1}$$

With E = 13,000 tons per sq.in., $\alpha = 11{\cdot}9 \times 10^{-6}$ per degree C., and $m = 3{\cdot}6$, the stress due to a difference in temperature of 1 deg. C. between frame and plate is obtained as 0·215 tons per sq. in. and for the actual temperature difference measured (3 deg. C.) it is 0·65 tons per sq.in. In actual fact it will be slightly lower because the frame is not perfectly rigid.

The notched tensile test piece of this material fractured in a brittle manner at −11 deg. C., which means that the transition temperature lies above −11 deg. C., and may be −9 deg. C., which is the temperature at which brittle fracture of the welded specimen occurred. It will be observed that the temperature at which brittle fracture occurred is well below the temperature of +30 deg. C. at which temperature the Izod impact value begins to fall away. The crack starting from the saw slots and running into the weld is shown in Figs. 17 and 18.

Third Test—Material 1

This specimen was identical with the one used in the second test, except that the frame and weld were stored at room temperature for three days before being inserted into the cooling chamber. This specimen cracked spontaneously when the plate material had reached the temperature of −7 deg. C. and the frame a temperature of −3 deg. C. The result of the test was identical with that obtained in the second test.

Fourth Test—Material 2

This specimen was identical with the specimens used in tests Nos. 2 and 3, except that the material was different and had a transition temperature below −43 deg. C. This specimen was cooled to −65 deg. C. but no cracking was observed. It had been stored for twenty-four hours after welding before being inserted into the cooling chamber. It was held at a temperature of −65 deg. C. for one-and-a-half hours. The lid of the cooling chamber was then removed and the specimen was heavily hammered whilst being kept at −65 deg. C., but no cracks occurred; the specimen was held at −65 deg. C. for a further five hours and then gradually rose in temperature over a further ten hours.

Fig. 16 (*below*). Second type of test plate with stiffeners. The hole and saw cuts are covered by the stiffener.

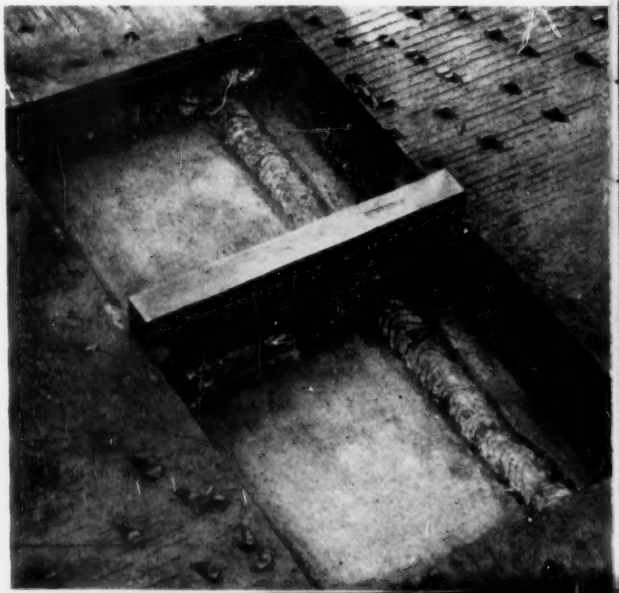


Fig. 17. Crack in Test 2, stiffened specimen, Material No. 1, commencing from saw cut and running into weld.

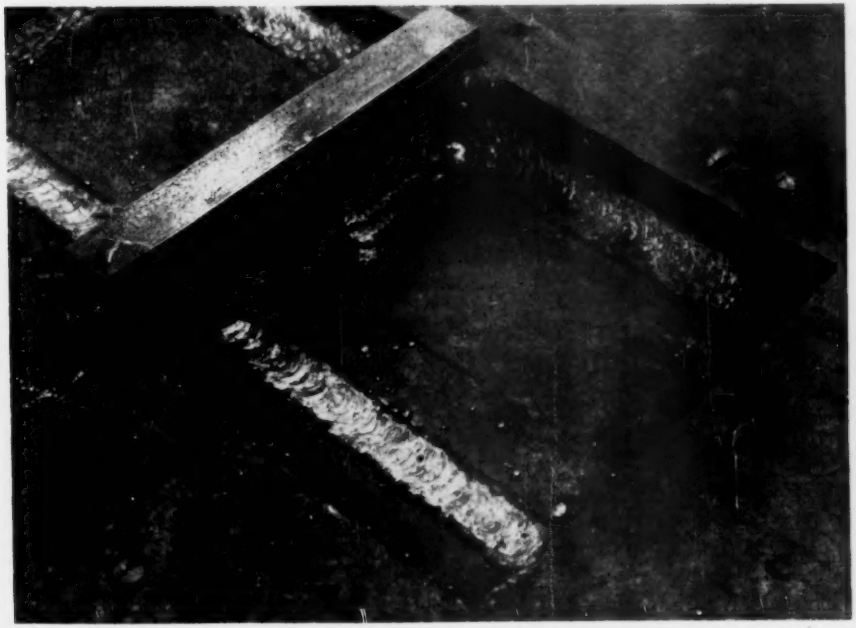


Fig. 18. Crack in test piece No. 2, stiffened specimen, Material No. 1.

Fifth Test—Material 3

This specimen was identical with those for tests Nos. 2, 3 and 4, except for the difference in the material and that all the welding was carried out with a preheat of 190 deg. C. The transition temperature of this material was below —64 deg. C. and the Izod value was still high at —60 deg. C. The specimen was cooled to —69 deg. C. after having been held for twenty-four hours at room temperature after welding. It was held at —69 deg. C. for one-and-a-half hours, when the lid was removed and the plate was heavily hammered, but no cracks occurred. The specimen was held for a further five hours at about —69 deg. C. and the temperature then gradually rose during a further eight hours.

DISCUSSION OF RESULTS

In considering the four similar tests with stiffened specimens first, it is found that tests Nos. 2 and 3, that is, those for which the coarse grained material (Material 1) was used, resulted in spontaneous brittle fracture evidently caused by residual stresses alone, whilst the other two test pieces made from fine grained steels (Materials 2 and 3) could be cooled to very much lower temperatures and could not be fractured even when severely hammered.

The spontaneous failures obtained in tests Nos. 2 and 3 are an exact reproduction of the well-known failures of some welded structures. It will be remembered that the failures of the Belgian and German bridges occurred without any apparent extraneous cause other than a considerable drop in the atmospheric temperature to some value in the neighbourhood of —20 deg. C. Similarly, the failure of specimens Nos. 2 and 3 occurred in consequence of a gradual reduction in temperature.

The magnitude and distribution of stresses in these four test pieces could not be ascertained. Even an approximate determination of the residual stresses would have required the test pieces to be destroyed by subdivision prior to their being inserted in the cooling chamber. Whatever the stresses were, there can be little doubt that the state of stress in all four specimens must have been closely similar since there was no difference in construction, welding procedure and sequence, except in test No. 5, where the welding was carried out with a preheat of 190 deg. C. The difference in composition and state of heat treatment cannot be conceived to have any influence on the state of residual stress introduced by welding. The fact that specimens Nos. 2 and 3 failed in exactly the same manner at temperatures within 2 deg. C. can surely be taken as an indication that there was very little, if any, difference between the stresses introduced in these two specimens. There is no reason to suppose that there should have been any more pronounced difference in the state of stress between, say, specimens No. 4 and No. 2 than there was between Nos. 2 and 3. The preheating of specimen 5 may have influenced the final state of residual stress, though it seems unlikely that an appreciable reduction in residual stresses can be obtained by preheating to 190 deg. C.

If the state of stress in the four specimens cannot have differed appreciably, only the difference in notch toughness of the three steels can account for the different behaviour of these four specimens. The temperatures at which the two specimens Nos. 2 and 3 failed and the lowest temperatures to which the other three specimens were cooled have been indicated in the diagrams giving the variation of Izod impact value with temperature (Figs. 10, 11 and 12). It will be observed that specimens Nos. 2 and 3 failed at temperatures at which the Izod impact value for Material 1 had dropped to approximately 8 and 10 ft. lb. respectively. At the lowest temperature to which specimens Nos. 4 and 5 were cooled the Izod impact values for Materials 2 and 3 were 11 and 65 ft. lb. respectively. Fracture of specimen No. 3 occurred at a temperature at which the Izod impact value of Material 1 was only slightly smaller than that obtaining for Material 2 at the lowest temperature to which specimen No. 4 was cooled and which did not fracture even when severely hammered at this temperature. It may be that specimen No. 4 was very near failure at —65 deg. C. and might have failed in the same way as specimens Nos. 2 and 3 had the temperature been reduced by, say, another 5 deg. or 10 deg. C. It may also be, and this seems likely, that the Izod test only gives an indication of the general tendency of a steel to lose notch toughness with decreasing temperature but is not sufficiently discriminating to allow a particular behaviour to be associated with actual numerical values of energy absorption.

From the test results obtained in this investigation it is impossible to say whether the Tipper test will give a more reliable indication of the behaviour to be expected at a certain temperature. It can only be said from these tests that a completely brittle fracture was obtained in the Tipper test with Material 1 at —11 deg. C., which is very near the temperature at which the welded specimens failed spontaneously, and that in the case of Material 3 a ductile failure was still obtained at —64 deg. C. and the welded specimen did not fail at —69 deg. C. The temperature reached in the Tipper test made from Material 2, —43 deg. C., was not low enough for any similar indication to be obtained. It may be that different results would have been obtained if the Tipper tests had been carried out with the full thickness of the material.

The small difference of 2 deg. C. between the failing temperature of specimen No. 2 and specimen No. 3 may not be significant, but it will be remembered that the plate used in specimen No. 3 was allowed to age for three days before cooling, and it may be that this accounts for the slightly higher failing temperature.

It does not appear, however, from these experiments that the low temperature notch toughness of the material is altered in the sense that the transition temperature range would be raised by the mere presence of residual stresses or that a very much less severe notch would be required for brittle fracture to occur at any given temperature than in an ordinary notch tensile or Izod test. The results of these experiments indicate that residual stresses do not embrittle the material. They may produce brittle fracture but only under very severe stress conditions, and then only at temperatures well below the transition temperature range of the steel.

In the earlier experiments carried out for the Admiralty Ship Welding Committee (*cf.* reference (2)), the specimens were not as severely restrained as those in the present series. The free length between restrained edges was greater, the width of the specimens was smaller, the plates were thinner and the edges of the plates at right angles to the butt weld were unrestrained. The restraining frame was also less rigid than that used in the present series of experiments. Even under these far less severe conditions of restraint, stresses equal to the yield point of the material could be shown to exist at least in the immediate vicinity of the weld. It was found, however, that these stresses, once the yield point was reached at any one point, could not be readily increased by superimposing external loading because of the plasticity of the material. If an external tensile load was superimposed, the stress at those points where the yield point was reached remained constant and only the residual stresses below the yield point could be diminished if they were compressive, or augmented if they were tensile. In

order to produce yielding all over the cross section a load of exactly the same magnitude would have been required as if no residual stresses had been present. Moreover, the stresses could not be increased anywhere beyond the yield point including those points where the residual stresses initially reached the yield point without first producing yield stress throughout the entire cross section. In order to produce stresses in excess of the yield point a load would, therefore, have been necessary of such magnitude as would have resulted in exactly the same stress in an initially stress-free specimen.

If one assumes that fracture, ductile or brittle, will not take place, at any rate until a stress considerably in excess of the yield point is reached, and if the stresses cannot be readily increased beyond the yield point, then the specimens, which were not notched, could not be expected to fail by brittle fracture under the influence of residual stresses alone, even when the temperature was reduced far below that at which brittle fracture could be produced in a notched tensile specimen. On the contrary, it was found that even at temperatures some 50 deg. C. below the transition temperature, plastic deformation at essentially constant stress took place when a specimen was intentionally cooled to a very much lower temperature than the frame, so that it would contract relative to the frame if it were not restrained. If this thermal contraction was prevented by restraint and if plastic deformation did not take place, additional thermal stresses would be superimposed on the specimen. In actual fact the specimen extended at constant stresses and the state of stress remained unchanged. This became obvious when, on allowing the specimens to warm up to room temperature, they were found to have slightly extended, resulting in a diminution of stress at room temperature.

In the present series of experiments care was taken that there was no appreciable difference in temperature between the frame and the specimen. A notch was introduced instead and it was thought that the residual stresses might be sufficiently intensified by the notch to initiate failure, which, if the temperature were low enough, would then rapidly progress through the entire cross section. This, however, did not happen in test No. 1, and it can only be assumed that despite the severe restraint the average residual stress value in the critical cross section was not large enough to produce fracture. It is just conceivable that the hot cracking at the weld and rewelding afforded some measure of stress relief in this specimen, so that after completion of welding the stresses were not as high as the severe conditions of restraint would normally be expected to produce. The rewelding of the hot cracks, however, took place under at least equally severe conditions of restraint as the original welding, so that this assumption does not appear very acceptable. It is more likely that the conditions of restraint, severe as they were, did not produce stresses high enough at a distance of 2 in. from the butt weld to produce failure.

The average stress in specimens Nos. 2 and 3 in which the free length was effectively halved by the stiffener, was probably much higher and was apparently sufficient to initiate and propagate the crack. It is also possible that the fillet welds connecting the stiffener produced triaxial stresses which aggravated the conditions at the bottom of the notch.

An interesting fundamental point is raised by these experiments. It has been suggested that brittle fracture cannot take place unless the stress reaches what is called the "Brittle Strength" of the material. This brittle strength is not very clearly defined. Some authors assume it to be independent of temperature,[4] others assume it to increase with decreasing temperature.[5] If this were the case, it would be very difficult indeed to explain the failures of specimens Nos. 2 and 3 and, incidentally, the failures of welded bridges and ships.

If it is assumed that more severe and more complex residual stresses than were present in specimens Nos. 2 to 5 are unlikely to be present in welded structures and that a saw cut represents as severe a constructional discontinuity as is ever likely to occur—unless there is a crack—one might draw the conclusion that residual stresses can produce spontaneous brittle fracture in welded structures only in the presence of severe discontinuities and at temperatures below the lowest temperature at which a notched tensile specimen would still fail in a ductile manner or an Izod impact test would still give a reasonably high value of energy absorption.

Even if high residual stresses and a severe notch effect are present, additional extraneous stresses due to non-uniform temperature changes, impact or static load may be required to produce failure, even if the temperature of the material is below the transition temperature. The fact that specimens Nos. 4 and 5 did not fail at low temperatures when severely hammered may be taken to indicate that the presence of residual stresses, even when a severe notch is present, will not facilitate brittle fracture under impact stresses as long as the temperature is above the transition range of the steel.

If there were a pronounced size effect other than that which arises from the increase in transition range with thickness, these tests on comparatively small assemblies might possibly give more favourable results than might be expected in full scale structures.

CONCLUSIONS

Although only five tests have been carried out and the character of the work was exploratory, the following tentative conclusions might be drawn:

(i) There has been no indication in the tests that residual stresses introduced by welding have an embrittling effect on material, i.e., making it liable to brittle failure at higher temperatures than those at which brittle fracture might occur in the absence of residual stresses.

(ii) Residual stresses of sufficient severity may produce spontaneous brittle fracture of welded assemblies without additional external stresses, but only if severe notches are present and if the temperature is below the lowest value at which a severely notched tensile test piece would break in a ductile manner, or at which the Izod impact strength would still be reasonably high.

(iii) In the tests there has been no indication that brittle fracture will occur whilst the structure is at a temperature above the Izod transition range of the material, even in the presence of severe notches and complex states of residual stress. These tests were carried out on small specimens and it is possible that conditions might be less favourable for larger assemblies if there is a size effect.

RECOMMENDATIONS FOR FURTHER WORK

In considering the possibility of brittle fracture arising from residual stresses, it is necessary to consider three variables:

(*a*) The state of residual stress introduced in consequence of welding. This depends on constructional detail and welding procedure and possibly the thickness of the material.

(*b*) The testing or service temperature of the material in relation to the transition temperature range, as determined by some specific form of notched test, and it must be remembered that different types of notched tests will give different transition temperature ranges.

(c) The severity of the constructional notch effect in relation to that used for the determination of the transition range and any possible size effect.

The five tests described in this report were exploratory in the sense that they showed qualitatively that residual stresses could produce brittle fracture and that the occurrence of brittle fracture was closely related to the transition range of the steel. It is impossible from these tests to answer any questions in a quantitative way, such as how severe and complex the state of stress must be to result in brittle fracture in a certain given material, with a definite transition range at a given temperature in the presence of certain notch effects. Considering the almost unlimited number of stress configurations which may arise in welded structures, it cannot be expected that the first question as to the severity of the state of stress required for brittle fracture can ever be answered in more than a general way. It should not be impossible, however, to devise some experimental set-up of which it could be said with absolute certainty that more severe stresses could not be expected to arise in any welded structure.

The other question of the relation between constructional notches and the transition temperature range of the material as determined by one definite type of notch test, could be profitably investigated further by using a similar experimental set-up as that used in the present investigation, which was very satisfactory. By varying the position and form of notch and the composition and thickness of the material, the relation between the transition range as determined by means of some definite form of notch test and the effect of various forms of constructional notches could be established.

Finally, it might be recommended to repeat the tests described in this report with three similar materials, but using considerably larger assemblies in order to determine whether there is a size effect or not.

ACKNOWLEDGMENTS

The author wishes to make grateful acknowledgment to Stewarts & Lloyds Ltd. for their generosity in providing all the material and facilities for this investigation to be carried out, to Mr. P. Whitaker, manager of Metallurgical Research, and to the staff of the Department of Research and Technical Development of Stewarts & Lloyds Ltd., Corby, for their collaboration.

REFERENCES

1. Weck, R., Stress Relieving of Welded Structures, *Welding*, 1949, October.

2. Weck, R., Transverse Contraction and Reaction Stresses in Butt Welded Mild Steel Plates, Report R.4, Admiralty Ship Welding Committee, *H.M.S.O.*, 1950.

3. Tipper, C. F. Elam, The Fracture of Mild Steel Plate, Report R.3, Admiralty Ship Welding Committee, *H.M.S.O.*, 1948.

4. *Cf.* for instance, Orowan, E., Fracture and Notch Brittleness in Ductile Metals, Report of Cambridge Conference on Brittle Fracture in Mild Steel, *B.I.S.R.A.*, 1945, p. 70.

5. *Cf.* Tipper, *Ibid* (4), p. 116.

R.84

ARGON-ARC WELDING OF COPPER

This report reviews the recent position of inert-gas-shielded arc welding of copper as given in the technical literature, together with some observations from work carried out in the B.W.R.A. laboratories. The report was issued to members of the Association in December, 1950.

By J. H. COLE, M.Sc.

INTRODUCTION

In inert-gas-shielded arc welding, usually referred to as argon-arc welding in Great Britain, an arc is drawn between a non-consumable electrode, usually pure tungsten, and the work piece. The electrode and the metal in the vicinity of the arc are shielded from the air by an atmosphere of some inert gas, helium or argon, which is introduced round the electrode. Any filler metal required is added independently.

Inert-gas-shielded arc welding was first applied industrially in 1942 for the welding of magnesium. Some time later it was also applied to stainless steel and aluminium where it has made rapid progress. More recently it has been applied to the less common metals with success. In many cases metals and alloys impossible to weld by other methods have been welded while in other cases the inert gas shielded welds have proved more economical and given better physical properties than welds produced by other methods.

In order to fusion weld a metal, heat must be put into the fusion zone more rapidly than it can be conducted away. Thus, because of its high thermal conductivity, the welding of copper is more difficult than most other metals. Preheating of the whole work piece will help.

With thin sections of copper an oxy-acetylene flame can put sufficient heat into the metal to melt it, but this becomes more difficult as the section thickness increases unless a lower melting point alloy is used to form the joint. This is not always acceptable because of the corrosion aspect.

The electric arc is the most intense source of heat available for welding. Metallic arc welding of copper has been tried with not very satisfactory results. Carbon-arc welding using a fluxed filler rod has been used with some success. Inert-gas-shielded arc welding appears to offer the best solution and has the following advantages over the other methods:

(*a*) Flux is not necessary.
(*b*) The concentrated heat source gives higher welding speeds.
(*c*) The heat-affected zone is narrower.
(*d*) Distortion is less.
(*e*) Reducing gases are absent.
(*f*) Welds with better physical properties are obtained.

REQUIREMENTS FOR ARGON-ARC WELDING OF COPPER

(*a*) *Power Supply*

The usual power recommended is direct current as supplied by the normal welding generator, with the electrode negative. Some investigators have welded pure copper and its alloys using alternating current and this is recommended for some of the alloys, *i.e.* aluminium bronze. Laboratory tests have indicated that using a $\frac{1}{1}$"leaky reactance" controlled transformer, at an open circuit voltage of 120 volts, it was not possible to prevent the arc from completely rectifying. Using a transformer having an open circuit voltage of 100 volts with an external choke control a fairly steady arc was maintained, but there was a relatively large D.C. component.

(*b*) *Superimposed High Frequency*

The use of a superimposed high voltage high-frequency current on the welding current is essential to maintain the arc when using alternating current and is preferred by many workers when using direct current. When using direct current the main purpose of the superimposed high frequency is to assist in starting the arc without touching the electrode on the work piece. Touch starting may cause contamination of the electrode together with an unsteady arc and loss of tungsten. Touch starting on a carbon block is less harmful but the pointed tip of the electrode is destroyed.

It has been found that the use of a thoriated tungsten rod together with a superimposed high frequency current gives much easier starting.

Any beneficial effects from the use of a superimposed high frequency current on D.C. power other than for ease of starting are rather doubtful.

(*c*) *Electrodes*

Pure tungsten electrodes have been used by practically all investigators. Laboratory tests have shown the advantage of a thoriated tungsten electrode (2 per cent. thoria) in giving easier starting. Carbon electrodes can be used for inert-gas-shielded arc welding and should be suitable for welding deoxidised copper.

The size of the electrode used depends on the current it is carrying (see Table I). When using D.C. the end of the electrode is ground to a point, with A.C. the tip of the electrode becomes molten and forms its own rounded end.

A consumable copper electrode has been tried, but with the equipment used it was not possible to maintain a steady arc. Recent development of the inert-gas-shielded metal arc welding indicate that the troubles associated with a consumable electrode have been at least partly overcome. The method has been used for welding aluminium bronze and may be applicable to the welding of copper.

(*d*) *Shielding Gas*

Most welding of copper in America has been carried out using helium as the shielding gas. Comparative tests using argon have given similar results. The arc voltage using helium is higher than that using argon, hence the energy input for the same current is greater. For the same reason the current in an argon arc is greater than that in a helium arc for the same current setting. This means that either the

Table 1. Typical Welding Conditions for Butt Welds in Copper, using Direct Current, Electrode Negative*

Material Thickness (in.)	Filler Rod Diameter (in.)	Electrode Diameter (in.)	Helium		Argon		No. of Passes	Pre-Heat Temp.	Edge Prep.
			Current D.C. Amps.	Gas Flow Cu.ft./hr.	Current D.C. Amps.	Gas Flow Cu.ft./hr.			
$\frac{1}{16}$	$\frac{1}{16}$	$\frac{1}{16}$–$\frac{1}{8}$	50-125	10-15	60-150	6-12	1	—	Square edge, close butt
$\frac{1}{8}$	$\frac{1}{8}$	$\frac{1}{16}$–$\frac{1}{8}$	125-225	12-20	140-280	7-15	1	—	Square edge, close butt
$\frac{1}{16}$	$\frac{1}{16}$–$\frac{1}{16}$	$\frac{1}{4}$	200-300	16-22	250-400	8-18	1	—	Square edge, open butt or 90 deg. bevel
$\frac{1}{4}$	$\frac{1}{16}$	$\frac{1}{16}$	250-350	14-30	300-475	16-25	1	600 deg. F. to 1000 deg. F.	90 deg. bevel, open butt
$\frac{3}{8}$	$\frac{1}{16}$	$\frac{5}{16}$	350-400	14	—	—	1	1000 deg. F.	90 deg. bevel, open butt
$\frac{1}{2}$	$\frac{1}{8}$	$\frac{1}{8}$–$\frac{1}{4}$	300-550	25-35	400-600	20-30	1	1000 deg. F.	90 deg. bevel, open butt
$\frac{1}{2}$	$\frac{1}{4}$	$\frac{5}{16}$	400	16	—	—	1	1000 deg. F.	90 deg. bevel, open butt or U-groove, open butt
$\frac{1}{2}$	—	$\frac{1}{4}$	275	—	—	—	4	—	90 deg. bevel, close butt

* From Mortimer, J. W., "Inert Gas Shielded Arc Welding Pure Copper," *The Welding Journal*, vol. 27, No. 9, pp. 715-722 (Sept. 1948); Sillifant, R. R., and Woolcott, W. A., "The Argonarc Welding Process," *Welding*, vol. 17, No. 11, pp. 493-506 (Nov. 1949); The American Welding Society *Welding Handbook*, Third Edition, pp. 756-757 (Jan. 1950).

welding speed will be less for the same current when using argon as against helium or to get the same welding speed a higher current is required when using argon. Both gases appear to be effective in protecting the electrode and the weld pool from oxidation, though a greater volume of helium is required.

Nitrogen, being inert towards copper, has also been tried. Low strength welds have resulted and there was considerable loss of tungsten from the electrode. This was transferred across the arc into the weld.

(e) Fluxes

One of the main advantages of inert-gas-shielded arc welding is that the inert gas protects the weld pool and the tip of the filler rod from oxidation and so renders the use of a flux unnecessary. Some investigators have used various brazing fluxes both on the plate and filler rod and found that they did not affect the weld quality but did prevent the formation of oxide on the parent metal in the vicinity of the weld. The area protected by the inert gas is fairly restricted.

(f) Filler Metal

When welding copper for chemical vessels it is usually essential that the composition of the weld deposit should not differ greatly from that of the parent plate, a deposit containing not less than 99 per cent. copper is usually acceptable.

The inert gas forms a protective atmosphere round the electrode and molten pool, but owing to the extreme agitation set up by the arc a certain small percentage of air is bound to be drawn in so some oxidation may take place. To counteract this possibility it is preferable to have some deoxidiser in the filler rod. The most effective additions for this purpose are silicon and tin which also give increased strength to the cast weld structure. Small additions of phosphorus alone do not appear to be satisfactory, the resultant welds being porous. Typical analysis of a suitable filler material is 0·25 per cent. silicon, 0·75 per cent. tin, 0·15 per cent. manganese and the remainder copper. This gives a weld deposit containing approximately 0·1 per cent. silicon, 0·4 per cent. tin and 0·1 per cent. manganese. Copper containing 0·25 per cent. silicon only has also given satisfactory results. A filler rod containing 1 per cent. silver and 0·05 per cent. phosphorus may be used on thin sheets. The strength of the weld metal is less than when using the other filler rod and the welds may be porous.

Where it is not necessary for the filler metal to be of the same composition as the parent metal other brazing materials may be used. An alloy containing 3 per cent. silicon, 1 per cent. manganese, remainder copper, has been used successfully as also has an alloy containing 80 per cent. copper, 15 per silver and 5 per cent. phosphorus. These brazing materials have a much lower melting point than copper and so offer the advantages of either less preheat or greater welding speed. Also, if the plate material is in the cold-worked condition, then the annealed zone adjacent to the weld may be narrower than if a copper filler rod were used. They are also helpful in joining thin sections to thick sections. Alloys containing zinc give off zinc fumes which make the brazing more difficult.

WELDING TECHNIQUE—BUTT WELDS

(a) Edge Preparation

With thin material under $\frac{1}{8}$ in. thick a square edge close butt joint with the addition of filler metal is satisfactory. Welds without the addition of filler rod are usually porous.

With material over $\frac{1}{8}$ in. thick the usual preparation is a close butt 90 deg. bevel, although up to $\frac{1}{4}$ in. thick a square edge open butt may be used. For thicknesses over $\frac{1}{2}$ in. a U preparation reduces the amount of filler metal required. The edges should be free from grease, dirt, etc.

(b) Position of Welding

Normally all welding is carried out in the downhand position with the plates horizontal or inclined 5 to 10 deg. up in the direction of welding.

Satisfactory welds have been made using a two-operator technique with the plate in the vertical position and an operator welding on each side.

(c) Backing

Molten copper is very fluid and so except when welding thin sheet a backing bar is normally used. The backing bar is made of copper and is either rigidly clamped or else tacked to the plates. The first run is melted into the backing bar to ensure complete penetration. After filling the weld groove the backing bar is chipped off and, if necessary, a sealing run laid on the back of the weld. Welding without a backing bar has also been carried out by leaving a $\frac{1}{8}$ in. land at the bottom of the bevel. After filling the groove the back was chipped out to sound metal and a sealing run laid down.

Welding from one side without a backing bar can be carried out by first putting down a small run and gradually filling the groove with additional runs. This requires much greater skill and leaves a rough and irregular surface at the back of the weld.

(d) Preheating

With thin sections preheating is not necessary, but with thicker sections it is essential for speed of operation and production of sound welds to preheat. Temperatures up to 1000 deg. F. (540 deg. C.) are used, the work being maintained at temperature during the welding. Smaller articles are heated in a small furnace while larger vessels, etc., are preheated using a gas torch. Increasing the preheat temperature either reduces the welding current required or increases the welding speed.

(e) Number of Passes

The number of passes required for a given plate thickness depends on the current used, the preheat temperature and the diameter of the filler rod.

Using a backing bar a $\frac{3}{8}$ in. thick plate may be welded in a single pass using a $\frac{1}{4}$ in. filler rod and 400 amps. direct current with a preheat temperature of 1000 deg. F. The fewer the number of passes the shorter the overall welding time, hence with high preheat temperatures it is more economical to weld with a single pass if possible. The annealing affect of one run on the previous run has no beneficial effect on the physical properties.

(f) Post Welding Treatment

There is no flux to be removed, but for use as chemical containers, etc., it is usual to grind the weld flush with the plate on those surfaces in contact with products of the process.

Hammering after welding may increase the strength of the joint in thin sections particularly if the weld is porous. Hammering is not necessary on thick sections and sound welds.

FILLET WELDS

Very little information has been published regarding fillet welds, corner joints, etc., although lap joints have been mentioned. The inert-gas-shielded arc welding process is readily adaptable for fillet welding and there does not appear to be any reason why satisfactory fillet welds cannot be made in copper.

The requirements for this type of joint may be such that a low melting point alloy could be used for the filler metal.

PLATE MATERIAL

Commercial copper sheet and plate is available in two distinct forms—tough pitch copper containing approximately 0·04 per cent. oxygen and deoxidised copper which is free from oxygen and usually contains traces of phosphorus.

Oxygen combines with copper to form cuprous oxide (Cu_2O) which forms a eutectic system with copper. The eutectic contains 3·5 per cent. cuprous oxide, equivalent to 0·3 per cent. oxygen and has a melting point of 1065 deg. C., which is 18 deg. C. less than that of pure copper.

In cast copper, the copper-copper oxide eutectic forms a brittle constituent between the copper grains, and, while having little effect on the strength causes a reduction in the ductility. After rolling to plate, the oxide is dispersed throughout the grains and has less effect on the physical properties.

During welding the copper is melted in the weld zone and any oxide present will again form a eutectic between the grains and give a structure with low ductility. Any air drawn into the arc atmosphere will increase the amount of oxide eutectic present.

The effect of copper oxide in the metal is further complicated by the presence of reducing gases such as hydrogen and carbon monoxide. The only source of hydrogen in the inert-gas-shielded arc apart from possible impurity in the shielding gas is from water vapour in the atmosphere which would dissociate in the arc. The hydrogen diffuses through the copper, reacts with the oxide to form steam which does not diffuse but remains entrapped and causes porosity.

Inert-gas-shielded arc welding has the advantage over other processes in that far less of the reducing gases, hydrogen and carbon monoxide are present to cause this porosity.

It would appear that under the influence of the arc certain elements or their oxides (e.g. phosphorus) partially volatilise and also cause porosity in the weld.

From the foregoing it is clear that to obtain sound welds with high strength and ductility the parent plate must be free from oxygen and the filler metal should contain a small percentage of non-volatile reducing elements.

The physical properties of welds in tough pitch copper are usually variable particularly as regards ductility as shown in bend tests. Higher welding speeds may assist in giving more consistent results but the ductility will always be lower than that obtained from sound welds in deoxidised copper.

The high preheat temperatures generally used will destroy all cold worked properties of the plate. Even without pre-heat the zone adjacent to the weld will be in the annealed condition so the strength of the weld cannot be greater than that of the annealed plate, and if the weld is sound, fracture in the tensile test will usually take place in this annealed zone.

Information on any change in the electrical resistance

Table II. Typical Results of Physical Tests on Inert Gas Shielded Arc Welds, as given by various authors

Material	Filler Metal Composition	Tensile Test		Free Bend per cent. Elongation on Outside	Guided Bend
		U.T.S. lb. per sq. in.	Elongation on 2 in.		
96 per cent. Cu, 3 per cent. Si, 1 per cent. Mn	Same	58,300	30-40	39	Free from cracks
Deoxodised copper	Copper plus ·25 per cent. Si ·75 per cent. Sn ·15 per cent. Mn	30,000 to 33,000	27-50	30	Free from cracks
Deoxodised copper	Copper plus ·25 per cent. Si	29,400	33-40	30	Free from cracks
Tough pitch copper	Copper plus ·25 per cent. Si ·75 per cent. Sn ·15 per cent. Mn	Up to 30,000*	Up to 36	Up to 38	Large proportion cracked

* Tensile test figures as low as 12,000 lb. per sq. in. have been quoted, but composition of filler metal was not given. Cracks usually form during bend tests.

Table III. Typical Welding Conditions for Butt Welds in Copper Alloy (96% Cu, 3% Si, 1% Mn) using Direct Current, Electrode Negative and Argon as Shielding Gas*

Material Thickness (in.)	Filler Rod Diameter (in.)	Electrode Diameter (in.)	Current D.C. Amps.	Gas Flow Cu.ft./hr.	No. of Passes	Edge Preparation
1/16	1/16-1/8	1/16-3/32	60-90	6	1	Square edge close butt
1/8	1/16-1/8	1/16-3/32	80-150	7-8	1	Square edge close butt
1/16	1/8-1/16	3/32-1/8	100-200	8-9	1	Square edge open butt or 90 deg. bevel
1/8	1/16-1/8	3/32-1/8	150-250	8-10	1	Square edge open butt or 90 deg. bevel close butt
1/4	1/16-1/8	1/8-3/16	290-300	—	5	90 deg. bevel close butt
1/2	1/4	1/4	150-300	8-10	2	90 deg. bevel close butt
1/4	1/4	1/4-5/16	250-400	9-11	3-4	90 deg. single or double bevel close butt
1	1/4	5/16-1/16	300-500	9-12	4-6	90 deg. single or double bevel close butt

* From Silliant, R. R., and Woolcott, W. A., "The Argonarc Welding Process," *Welding*, vol. 17, No. 11, pp. 493-506 (Nov. 1949).

across inert-gas-shielded arc welds is very sparse, but one set of results showed that the resistance across the weld when all reinforcement had been removed was 15 per cent. greater than that of the plate. Any reinforcement will, of course, decrease the resistance.

COPPER ALLOYS

Large tanks, made of an alloy of approximately 96 per cent. copper, 3 per cent. silicon and 1 per cent. manganese, have been welded using a filler metal of the same composition. Typical welding conditions are shown in Table III.

Aluminium bronze has been satisfactorily welded and has also been welded to other copper alloys to give a hard surface. It has been shown that this alloy can be readily welded using the inert-gas-shielded metal arc welding.

Examples have been quoted in the literature showing the welding of most of the normal copper alloys even including a 70-30 brass containing 2½ per cent. lead. Alloys containing zinc give off zinc fumes which may upset the welder, but do not greatly affect the welding process. The colour match is not always perfect owing to small differences in composition and structure between the parent metal and weld deposit.

CONCLUSIONS

Deoxidised copper sheet or plate can be welded by this method, the welds having the strength and ductility of the fully annealed parent metal. Oxygen-containing (i.e., tough pitch) copper sheet or plate can be welded, but the physical properties of the joint are variable. Alloys of copper can also be satisfactorily welded.

Selected List of References

Herbst, H. J. Production Applications for Inert Gas Shielded Arc Welding, *The Welding Journal*, vol. 26, No. 5, pp. 410-418 (May 1947).

Gurklis, P. J. Silicon Bronze Welding Problems Overcome through Inert-Arc Method, *Materials and Methods*, vol. 27, No. 2, pp. 66-69 (Feb. 1948).

Jewell, S. V. Copper Tanks Welded by Inert-Arc, *The Welding Engineer*, vol. 33, No. 7, pp. 33-37 (July, 1948).

Mortimer, J. W. Inert-Gas-Shielded Arc Welding Pure Copper, *The Welding Journal*, vol. 27, No. 9, pp. 715-722 (Sept. 1948).

Koopman, K. H. Hard Facing with Inert-Gas-Arc Welding, *The Welding Journal*, vol. 28, No. 1, pp. 46-52 (Jan. 1949).

Huff, H. A., and Kugler, A. N. Practical Aspects of Inert-Gas Welding, *The Welding Journal*, vol. 28, No. 2, pp. 128-140 (Feb. 1949).

Silliant, R. R., and Woolcott, W. A. The Argonarc Welding Process, *Welding*, vol. xvii, Nos. 10 and 11, pp. 430-441, 493-506 (Oct. and Nov. 1949).

Kelley, J. D. Heavy Gauge Copper Welded by Inert-Gas-Shielded Arc Process, *Materials and Methods*, vol. 31, No. 2, pp. 57-58 (Feb. 1950).

Muller, A., Gibson, G. J., and Roper, E. H. The Aircomatic Welding Process, *The Welding Journal*, vol. 29, No. 6, pp. 458-482 (June, 1950).

Printed in England by Lawrence Press Ltd., Henry Road, New Barnet, Herts